phys. stat. sol. (a) **76**, 437 (1983)

Subject classification: 1.5 and 10.2; 22; 22.1.1; 22.6

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Accommodation of Misfit in Heterostructures with Continuous and Island Films

By

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An equation is derived for the equilibrium dislocation configuration in a slip plane with a stress distribution depending on a single coordinate. The equation allows to determine the critical film thickness at which dislocations are first formed in various heterostructures. Critical thicknesses are discussed for systems with a continuous film, with a semi-infinite film, bounded by a straight edge, and with a series of parallel strips on the substrate surface. The theoretical data are in agreement with the observed ones for continuous films. They are confirmed by experimental results for the heterostructure with a discontinuous film. The peculiarities of stable and unstable equilibrium dislocations are discussed for the mentioned heterostructures.

Найдено уравнение равновесной дислокационной конфигурации для случая, когда сдвиговое напряжение в плоскости скольжения зависит от одной переменной. Данное уравнение позволяет определять критическую толщину пленки при возникновении дислокаций в различных структурах. Определена критическая толщина в случаях сплошной пленки, полубесконечной пленки, ограниченной одиночным прямолинейным краем и для группы параллельных полосок пленки. Теоретические результаты согласуются с известными для структур со сплошными пленками и подтверждены экспериментально для гетероструктуры с несплошной пленкой. Для различных гетероструктур обсуждаются особенности устойчивых и неустойчивых равновесных дислокации.

1. Introduction

Misfit dislocations are widespread defects in heterostructures with continuous films. In heteroepitaxial structures they arise due to the difference in lattice parameter f of juxtaposed materials [1, 2]. In systems with amorphous layers f is equal to a relative size misfit of a film and substrate taken along the interface [3 to 5]. The critical conditions for the formation of the misfit dislocations have been investigated in all detail [1 to 8].

In addition to the misfit dislocations in heterostructures with island films the dislocations caused by an inhomogeneous stress distribution near film edges appear. In the literature they are referred to as emitter edge [9], stress-jumping [10], boundary [11], and other dislocations. The glide mechanism of these dislocations [11 to 13] as well as their stable equilibrium dislocation configurations (EDCs) [8, 13 to 15] have been investigated theoretically and experimentally. However, the existence of critical configurations for their formation has been proved only experimentally [16 to 18] without any theoretical considerations.

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In the present paper a more general expression than the known one is derived for the critical film thickness of heterostructures with continuous and island films. Theoretical and experimental results are compared for discontinuous films such as Si_3N_4 on Ge(111) substrates.

2. Theoretical Considerations and Results

2.1 Derivation of the main equations

If Peierls stresses are neglected, the expression

$$R = W/(b\tau_b) \tag{1}$$

is known to hold true at the points of EDCs [8]. Here R is the radius of the dislocation curvature, W the dislocation energy per unit length, b the magnitude of the Burgers vector $\boldsymbol{b}$, τ_b the resolved shear stress along $\boldsymbol{b}$ in the slip plane. If $y(x)$ is the equation describing the equilibrium shape of a glide dislocation, then taking into account the dependence of R on $y'(x)$ and $y''(x)$, one obtains

$$y'' = \frac{\pm[1 + (y')^2]^{3/2}\, b\tau_b}{W}. \tag{2}$$

If $x0y$ is the slip plane, then in general τ_b is a function of x and y. But in some cases the dependence on one of the variables vanishes. This is the case for heterostructures with continuous films and for heterostructures with discontinuous films if the trace of the slip plane with the surface is parallel to the film edge. The analysis will now be carried out on the assumption that τ_b depends only on x and that W remains constant. Not restricting the general character of our considerations one can assume that the curve $y(x)$ passes through the origin of the coordinate system. Using the initial conditions $y(0) = 0$ and $y'(0) = k$ one obtains

$$y(x) = \pm \int\limits_0^x \left\{\left[\left(\frac{b}{W}\right) \int\limits_0^\xi \tau_b(\chi)\, d\chi \pm k(k^2 + 1)^{-1/2}\right]^{-2} - 1\right\}^{-1/2} d\xi \tag{3}$$

by separating the variables. For further analysis we need the following function of $y'(x)$:

$$[y'(x)]^{-2} = \left[\left(\frac{b}{W}\right) \int\limits_0^x \tau_b(x)\, dx \pm k(k^2 + 1)^{-1/2}\right]^{-2} - 1. \tag{4}$$

Two curves (3) corresponding to the different signs before the term $k(k^2 + 1)^{-1/2}$ meet the initial conditions. In the vicinity of zero they are located on each side of the common tangent. The curve given by (3) has a tangent parallel to $0y$ at the point x^*, $y(x^*)$ if $[y'(x^*)]^{-2} = 0$, i.e.

$$b \left|\int_0^{x^*} \tau_b(x)\, dx \pm Wk(k^2 + 1)^{-1/2}\right| = W. \tag{5}$$

For the initial conditions $y(0) = 0$, $y'(0) = 0$

$$b \left|\int_0^{x^*} \tau_b(x)\, dx\right| = W. \tag{6}$$

If $\tau_b(x) = \tau_b(-x)$ and the level of stress is rather large, then the curve satisfying (6) is a closed loop with the largest size $2x^*$ along $0x$ and $2y(x^*)$ along $0y$. The dislocation shape satisfying (5) is closed, when the values $y(x_1^*)$ and $y(x_2^*)$ are equal at $x_1^* < 0$ and $x_2^* > 0$. If stresses are rather large one can find k in order to get such EDC.

Equation (5) is derived for the first time. A particular case (6) was considered earlier [2 to 4] where the forces acting on dislocation segments were analysed.

Using (5) and (6) we shall first consider the problem of the critical conditions for heterostuctures with continuous films discussed also elsewhere [1 to 4] and then we shall be concerned with the original results for structures with discontinuous films. Furthermore the elastic constants of film and substrate will be assumed to be the same.

2.2 Formation of equilibrium configurations in systems with continuous films

Here the coordinate planes for the film ($x_d 0_d y_d$) and for the substrate ($x_s 0_s y_s$) coincide with the slip plane (Fig. 1), and the axis of ordinates is the intersection of the slip plane with the free surface of the film ($0_d y_d$) or of the substrate ($0_s y_s$). We shall discuss the problem for dislocations, reaching the free surface of the film or of the substrate since this is usually the case. If the film thickness h is much less than the substrate thickness H, then in the film the stress τ_{bd} and the deformation ε_d are independent of the distance from the surface. Therefore, the equilibrium configuration $y_d(x_d)$ is a circular arc reaching the point 0. The critical film thickness h_d^c at which misfit dislocations are first formed by $y_d(x_d)$ is determined by the condition $[y'(x)]^{-2} = 0$ at $x = h/\sin\theta$, i.e. the interface is the plane tangent to 0_dMN arc (Fig. 1). Here θ is the angle between the slip plane and the interface. The value $x^* = h/\sin\theta$ must meet (5) or (6) depending on the magnitude of $y_d'(0)$.

In the substrate the stresses τ_{bs} vary from the maximum value $\tau_{bs\,(max)}$ near the interface $x_s = H/\sin\theta$ to the value $-\tau_{bs\,(max)}/2$ at the free surface, being zero along the neutral surface $x_s = H/(3\sin\theta)$. Since the signs of the stresses τ_b are opposite in the film and in the substrate near the interface one can assume $\tau_b = 0$ at the interface, i.e. $t_{bd}(h/\sin\theta) = \tau_{bs}(H/\sin\theta) = 0$. Thus the straight line $x = h/\sin\theta$ satisfies (2) and when the film thickness h_d^c is critical, the sections 0_dM and NM (Fig. 1) glide to the opposite directions and pull the equilibrium misfit dislocation M_1M_2. This dislocation is in a stable equilibrium state as its small displacement from the interface into the film or substrate gives rise to an opposing force. With the exception of M_1 and M_2 points the 0_1M_1 and N_1M_2 gliding sections are in an indifferent equilibrium state. As the

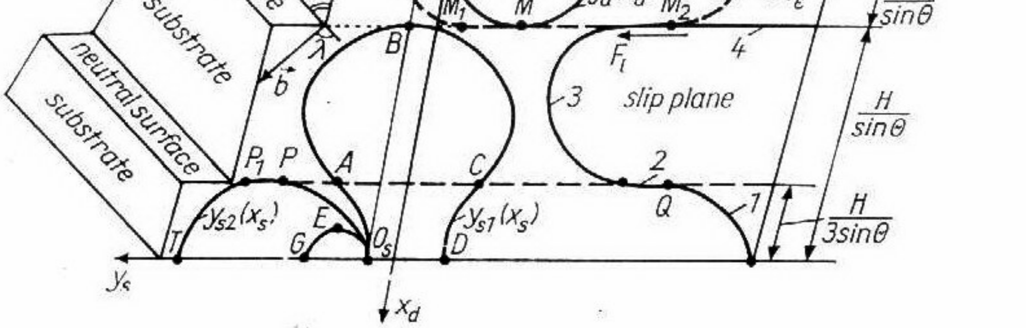

Fig. 1. Scheme of equilibrium configurations of a heterosystem with a continuous film. $x_d 0_d y_d$ coordinate system for film dislocations, $x_s 0_s y_s$ for substrate ones; M_1M_2 and 4 misfit dislocations in the interface, PP_1 and 2 straight discolation parts in the neutral surface

function $\tau_b(x)$ is discontinuous at $x = h/\sin\theta$, the dislocation is in a nonequilibrium state at the points M_1 and M_2.

Allowing for the assumption $W = \text{const}$ (see Section 2.1) one can see that $\int_0^{x^*} \tau_b(x)\,dx$ is the force acting on the gliding section F_ε (Fig. 1) and W is the force of the line tension (F_l) of M_1M_2. Since at the critical film thickness $F_\varepsilon = F_l$, the formation of M_1M_2 is determined by (6), and NM and 0_dM are normal to the trace of the slip plane on the surface and 0_dMN is a semicircular dislocation. If the length of 0_dMN were supposed to be smaller than πR, then the pulled misfit dislocation M_1M_2 would be in a metastable equilibrium state. Using (5) one can show that the energy gain equals $Wk(k^2+1)^{-1/2}$ when the metastable dislocation passes through the free surface.

Note that if $h > h_d^c$, the radius of an unstable equilibrium half-loop is smaller than $h/\sin\theta$, and the equilibrium misfit dislocation is formed by nonequilibrium gliding sections.

In [2] Matthews has compared the values h^c, determined from the equilibrium of forces acting on the dislocation (see (15) [2]) with those determined from the energetical analysis of the dislocation arc generated at the film surface (see (24) and (25) [2]). The equations correlate well for the semicircular arc, which is in agreement with our results.

In the substrate $\tau_{bs}(H/\sin\theta) = \tau_{bs}[H/(3\sin\theta)] = 0$. Thus, there are two critica film thicknesses [3, 4]: h_{s1}^c — for the formation of misfit dislocations by EDCs $y_{s1}(x_s)$ with $[y'(H/\sin\theta)]^{-2} = 0$ and $[y'(H/(3\sin\theta))]^{-2} \neq 0$ (Fig. 1) and h_{s2}^c — for the formation of straight dislocations in the neutral surface by $y_{s2}(x_s)$ with $[y'(H/(3\sin\theta))]^{-2} = 0$. The stable misfit dislocation $x = H/\sin\theta$ is formed at the film thickness h_{s1}^c when the sections 0_sAB and DCB (Fig. 1) are in the indifferent state and glide to opposite directions. The peculiarities of the process are similar to those taking place during formation of the misfit dislocations by the film dislocations.

When $h \gg h_{s1}^c$ the equilibrium half-loop 0_sEG (Fig. 1) contains the point E with the abscissa $x^* < H/(3\sin\theta)$ satisfying $[y'(x^*)]^{-2} = 0$. One can show that $y(x^*)$ tends to infinity when x^* tends to $H/(3\sin\theta)$. The function $\tau_b(x)$ is continuous in the vicinity of the neutral surface, thus an accurate analysis of the formation of the straight dislocation segments is possible. But here we use the previously discussed simplified model: in the indifferent equilibrium state the sections 0_sP and P_1T (Fig. 1) glide in opposite senses and form the straight stable dislocation part PP_1.

Note that at h_{s2}^c the dislocation configurations with sections 1 and 3 (Fig. 1) pulling the stable straight part 2 in the neutral surface and the stable misfit dislocation 4 are possible, too [3, 4]. Section 1 is in an indifferent equilibrium state. Section 3 and the dislocation piece in the vicinity of point Q are nonequilibrium. Just as the $y_d(x_d)$ arc, the equilibrium configurations $y_{s1}(x_s)$ and $y_{s2}(x_s)$ and the equilibrium section 1 make right angles with the ordinate axes.

To derive the equations for the discussed critical thicknesses we assume $\varepsilon_d = f$. Thus $\tau_{bd} = 2G(1+\nu)\,f\cos\lambda\sin\theta/(1-\nu)$ in the film [8]. $\tau_{bs(max)} = (1+\nu) \times 8Gfh\cos\lambda\sin\theta/[H(1-\nu)]$ in the substrate near the interface, and $\tau_{bs}(x) = 3x\tau_{bs(max)}\sin\theta/(2H) - \tau_{bs(max)}/2$ at any point of the slip plane. Here G is the shear modulus, ν the Poisson ratio, λ the angle between $\boldsymbol{b}$, and the projection of $0x$ on the interface. Using these equations and (6) one gets

$$h_d^c = h_{s1}^c = W(1-\nu)/[2G(1+\nu)\,bf\cos\lambda]\,, \tag{7}$$

$$h_{s2}^c = 3W(1-\nu)/[2G(1+\nu)\,bf\cos\lambda]\,. \tag{8}$$

Equations (7) and (8) are in agreement with the equations derived in [3, 4].

2.3 Equilibrium dislocations in substrates with discontinuous films

The slip plane will now be parallel to the interface. The coordinate plane $x0y$ coincides with the slip plane in such a way that $0y$ may be the projection of the film edge (Fig. 2a, 3a).

2.3.1 Heterostructures with semi-infinite film bounded by a straight edge

In this case the stresses $\tau_b(x) = \tau_b(-x)$ (Fig. 2b). Therefore, we shall consider EDCs satisfying $y'(0) = 0$. Later this assumption will be explained. If at some h^o $\tau_s^o(x)$ is known, then at any h the stress distribution $\tau_b(x)$ is expected to be

$$\tau_b(x) = \frac{\tau_b^o(x)\, h}{h^o}. \tag{9}$$

The analysis of the dependence of the energy on R, considered in [8], gives the direct relation between R and τ_b for the stable equilibrium state and the inverse relation for the unstable one. By using (6) and (9) one can find the finite x^* at a rather large h and get closed EDC $y(x)$ (Fig. 2a) which is in the unstable state as far as x^* decreases with the growth of h.

The critical film thickness at which the closed unstable loops appear under the film edge is determined by $x^* = \infty$. Thus it is (see (6) and (9))

$$h^c = \frac{W h^o}{\left| b \int\limits_0^\infty \tau_b^o(x)\, dx \right|}. \tag{10}$$

It remains to study the dependence of h^c on f. If the film edge effect (Fig. 2e) is substituted by a concentrated line force $\sigma_d h$ tangential to the interface and normal to film edge, then the shear stresses [19, 20]

$$\tau_{zx} = -\frac{(2\sigma_d h/\pi)\, x^2 z_{in}}{(x^2 + z_{in}^2)^2}, \tag{11}$$

where σ_d is the normal stress in the film, z_{in} the distance between the interface and the slip plane. Equation (11) turns to 0 at $0y$ which is at variance with the experimental results [13] and accurate numerical calculations [21]. Even so, the distortion is assumed to make small alterations in the integral. Using (6), (11), and expressions $\tau_b = \tau_{zx} \cos\lambda$ and

$$\sigma_d = \frac{2Gf(1+\nu)}{(1-\nu)}, \tag{12}$$

one gets

$$h^c = \frac{W(1-\nu)}{G(1+\nu)\, bf \cos\lambda} \tag{13}$$

for any z_{in}. The dependence of h^c on z_{in} is determined by $W(z_{in})$ and the inverse dependence of h^c on f is explained to be the direct relation between τ_b and f for any fixed x. It is seen that h^c determined by (13) is twice as high as that determined by (7) at the same parameters of heterostructures.

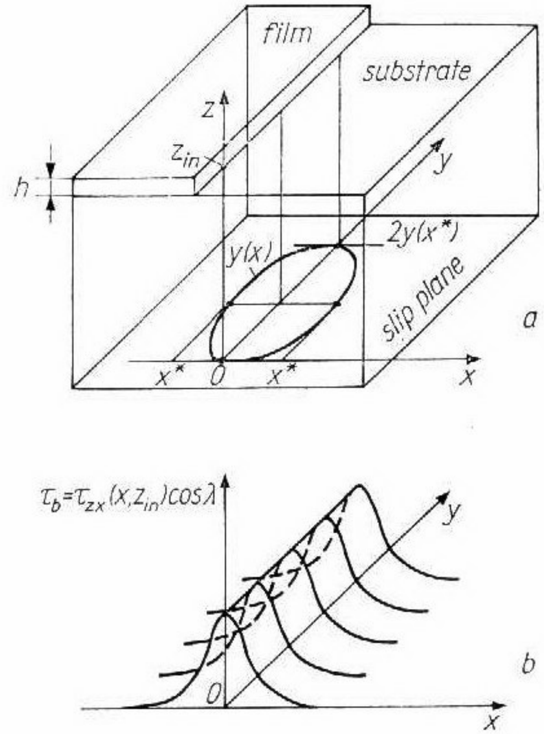

Fig. 2

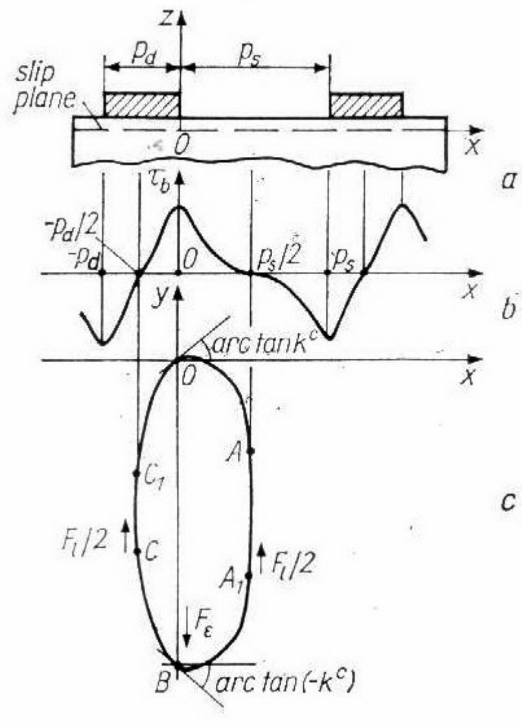

Fig. 3

Fig. 2. Heterostructure with a semi-infinite film. a) The unstable equilibrium substrate dislocation $y(x)$ gliding under the straight edge of the film; b) distribution of the stress component $\tau_b(x)$ in the slip plane

Fig. 3. Heterostructure with infinitely long bands of a film and open substrate surface. a) Scheme of bands, b) distribution of the stress component $\tau_b(x)$ in the slip plane parallel to the interface, c) equilibrium dislocation loop with gliding sections C_1OA and A_1BC which pull straight dislocation parts AA_1 and CC_1

2.3.2 EDCs in substrate with "zebra film"

The heterosystem will now be a series of parallel film strips separated by open substrate surface strips (Fig. 3a). $0y$ is the projection of any film edge. If the width of the film strip (p_d) is not equal to the one of the substrate strip (p_s), then taking account of the stress distribution (Fig. 3b) one can write

$$\left|\int_0^{-p_d/2} \tau_b(x)\,dx\right| \neq \left|\int_0^{p_s/2} \tau_b(x)\,dx\right| . \tag{14}$$

For a rather large h there are x_d^* and x_s^* ($-p_d/2 < x_d^* < 0 < x_s^* < p_s/2$) satisfying $[y'(x_d^*)]^{-2} = [y'(x_s^*)]^{-2} = 0$ and the closed unstable dislocations loops exist within the range of $-p_d/2 < x < p_s/2$. The loops will expand and in a nonequilibrium state they will form the straight dislocation parts along $x = -p_d/2$ and $x = p_s/2$ lines. As $\tau_b(-p_d/2) = \tau_b(p_s/2) = 0$, the straight parts are in equilibrium. Their states are stable just like the states of parts 2 and 4 in Fig. 1. When $x_d^* \to -p_d/2$ and $x_s^* \to p_s/2$, the value h tends to the critical thickness h^c which is determined by the conditions $[y'(-p_d/2)]^{-2} = [y'(p_s/2)]^{-2} = 0$. So far as $p_s \neq p_d$, the demand for $k = y'(0) = 0$ is not sound. Using (5) and (9) one can show that two sets of simultaneous equations

$$\left(\frac{hb}{h^0}\right)\left|\int_0^{-p_d/2} \tau_b^0(x)\,dx\right| \pm Wk(k^2+1)^{-1/2} = W ,$$

$$\left(\frac{hb}{h^0}\right)\left|\int\limits_0^{p_s/2} \tau_b^0(x)\,dx\right| \mp Wk(k^2+1)^{-1/2} = W \tag{15}$$

are valid for the critical values of h and k. In the sets the upper or lower signs before the terms $Wk(k^2+1)^{-1/2}$ should be taken in pairs. The critical value of h is determined by

$$h^c = \frac{2Wh^0}{b\left(\left|\int\limits_0^{-p_d/2} \tau_b^0(x)\,dx\right| + \left|\int\limits_0^{p_s/2} \tau_b^0(x)\,dx\right|\right)} \tag{16}$$

or any set of (15). The sets (15) have two roots of k:

$$k = \pm k^c = \pm\left[(2Wh^0)^2/\left(h^c b\left|\int\limits_0^{-p_d/2} \tau_b^0(x)\,dx\right| - h^c b\left|\int\limits_0^{p_s/2} \tau_b^0(x)\,dx\right|\right)^2 - 1\right]^{-1/2}. \tag{17}$$

According to Section 2.1 conditions $y(0) = 0$ and $y'(0) = \pm k$ give four EDCs. One can show that only two curves have tangents with both $x = -p_d/2$ and $x = p_s/2$ lines. The section of one of them is shown in Fig. 3c. It is $C_1$0A satisfying $y(0) = 0$ and $y'(0) = k^c$. If the origin of the coordinates is supposed to be at the point B, the curve A_1BC will satisfy the conditions $y(0) = 0$ and $y'(0) = -k^c$. Just like the formation of straight dislocations at the substrate neutral surface (see Section 2.2), their formation in the middle of the strips is discussed within the framework of the simplified model. The loops consist of the two straight dislocation parts AA_1 and CC_1 and of the two sections AOC_1 and A_1BC (Fig. 3c). The sections can glide in opposite directions and elongate the straight parts. The sum of the left-hand sides of (15) is the force $F_\varepsilon = (hb/h^0)\int\limits_{p_d/2}^{p_s/2} \tau_b^0(x)\,dx$ which acts on the gliding sections (e.g. A_1BC in Fig. 3c) and the sum of the right-hand sides of (15) is the resultant force of the line tension of parts AA_1 and CC_1 ($F_e = 2W$). As the sums are equal, the force equation $F_l = F_\varepsilon$ takes place and the sections A_1BC and AOC_1 are in an indifferent equilibrium state. If the straight dislocation parts appeared at $h < h^c$ they would be in the metastable state and the annihilation of neighbouring parts would decrease the energy of the system ($F_\varepsilon < F_l$).

It is evident from (17) that at $p_d = p_s = p$ the value of k^c is equal to 0 and EDC is symmetrical with respect to $0y$. As it is true for $p \to \infty$, the assumption $k^c = 0$ taken in Section (2.3.1 for the estimation of h^c in the case of a continuous film with a single straight edge is clear.

So we have determined the critical film thickness for the heterostructures with "zebra films" by starting the formation of stable straight dislocations in the strips. The existence of closed stable EDC is impossible in heterostructures with infinitely long bands. They appear in the case of rectangular islands [8, 13] or more complicated patterns of the discontinuous film. In heterostructures with "zebra film" unclosed stable dislocations can be formed by initial dislocations of the parent substrate crystal at $h < h^c$. But this question will be discussed elsewhere.

2.3.3 *Estimation of critical discontinuous film thickness*

The results of [13, 14] were used for an experimental verification of the theory. Dislocations appeared in Ge substrate under rectangular $Si_3N_4 + SiO_2$ islands (Fig. 4) [14]. We used (16) as far as islands were elongated. Single straight dislocations and their groups (see arrows in Fig. 4) are shifted from the band middles due to a small

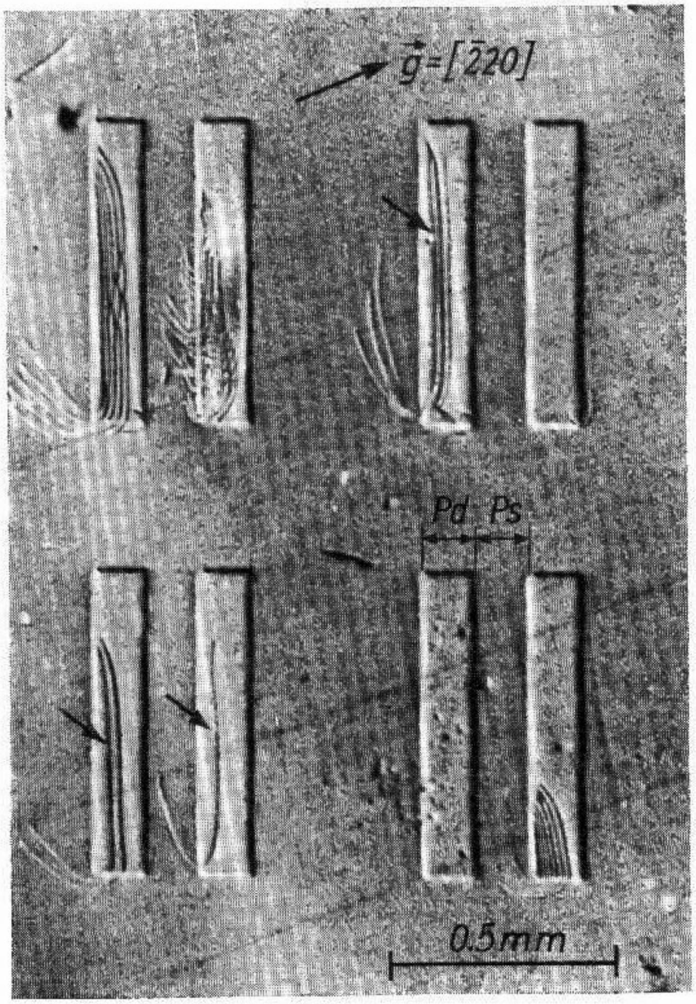

Fig. 4. X-ray topograph of Ge–Si_3N_4 + SiO_2 system with dislocations gliding in planes (111) roughly parallel to the interface [14]. The single straight dislocation part and groups of the parts are denoted by arrows

misorientation between (111) and the interface. (This effect allowed the normal stresses to be calculated [13]).

To estimate h^c by (16) we used $W = 1.1Gb^2[\ln(r/b) + 1]/(4\pi)$ [8], $G = 5.4 \times 10^{10}\ \mathrm{N/m^2}$, $b = 0.4$ nm, the middle depth of the dislocation position $r = 9\ \mu$m and the experimental curve $\tau_b^0(x)$ given in [13] for $h^0 = 0.12\ \mu$m, $p_d = 120$ and $p_s = 104\ \mu$m (see Fig. 4). One can get $|\int_{p_d/2}^{0} \tau_b^0(x)\,\mathrm{d}x| + |\int_{p_s/2}^{0} \tau_b^0(x)\,\mathrm{d}x| = 4 \times 10^{-3}$ N/m and $h^c = 0.15\ \mu$m, which is about h^0. According to Fig. 4 and the data of [13] the dislocations have not appeared under some islands. Thus, the used film thickness is really close to the critical one.

3. Conclusion

Though the expressions obtained are approximate, they permit us to understand the main features of the dislocation structure formation. If we had an accurate equation for τ_{zx}, we should have derived the more correct dependence of the critical film thickness on the slip plane depth z_{in}. To study the nonequilibrium configurations, the metastable dislocations at the precritical film thickness, and other peculiarities of EDCs, it is necessary to carry out "in situ" experiments.

Acknowledgements

The author wishes to thank Dr. S. I. Stenin, Dr. K. K. Ziling, S. M. Pintus, and A. G. Noskov for the fruitful discussion and L. P. Kibireva for the help at some stages.

References

[1] F. S. Frank and J. H. Van der Merwe, Proc. Roy. Soc., Ser. A **198**, 216 (1949).
[2] J. W. Matthews, J. Vacuum Sci. Technol. **12**, 126 (1975).
[3] E. M. Trukhanov and S. I. Stenin, phys. stat. sol. (a) **66**, 125 (1981).
[4] E. M. Trukhanov and S. I. Stenin, Fiz. i Khim. Obrabotki Materialov **5**, 148 (1979).
[5] E. M. Trukhanov and L. Zsoldos, phys. stat. sol. (a) **66**, 157 (1981).
[6] A. K. Gutakovskii, B. G. Zaharov, S. I. Stenin, and V. M. Ustinov, Izv. Akad. Nauk SSSR, Ser. fiz. **41**, 2301 (1977).
[7] G. A. Rozgonyi, P. M. Petroff, and M. B. Panish, J. Cryst. Growth **27**, 106 (1974).
[8] E. M. Trukhanov and S. I. Stenin, phys. stat. sol. (a) **66**, 591 (1981).
[9] G. H. Schwuttke and K. Howard, J. appl. Phys. **39**, 1581 (1968).

[10] G. H. Schwuttke, Microelectronics and Reliability 9, 397 (1970).
[11] A. Cerutti and C. Ghezzi, phys. stat. sol. (a) **17**, 237 (1973).
[12] I. A. Blech, E. S. Meieran, and H. Sello, Appl. Phys. Letters 7, 176 (1965).
[13] E. M. Trukhanov, S. I. Stenin, and A. G. Noskov, phys. stat. sol. (a) **53**, 433 (1979).
[14] E. M. Trukhanov, S. I. Stenin, and A. G. Noskov, phys. stat. sol. (a) **56**, **443** (1979).
[15] Svoistva struktur metall-dielektrik-poluprovodnik, Ed. A. V. Rzhanov, Izd. Nauka, Moskva 1976 (p. 222).
[16] A. Bohg and A. K. Gaind, Appl. Phys. Letters **33**, 895 (1978).
[17] V. M. Tuluevskii, U. Ya. Ertelis, and I. A. Feltynsh, Elektronnaya Tekh., Ser. 2, No. 8, 76 (1974).
[18] J. Tamaki, S. Isomae, S. Mizuo, and H. Higuchi, J. Electrochem. Soc. **128**, 644 (1981).
[19] S. M. Hu, J. appl. Phys. **50**, 4661 (1979).
[20] S. Isomae, J. appl. Phys. **52**, 2782 (1981).
[21] T. G. Beleytcheva, Zh. prikl. Mekh. i tekh. Fiz. No. 5, 135 (1979).

(Received July 14, 1982)